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EFFECT OF STRESS ON CREEP AT HIGH TEMPERATURES

THIRTY-SEVENTH TECHNICAL REPORT

BY

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INSTITUTE OF ENGINEERING RESEARCH
UNIVERSITY OF CALIFORNIA, BERKELEY

September 1, 1954

Office of Naval Research
Department of the Navy
Washington 25, D. C.

ATTENTION: Mr. Julius Harwood

Dear Sir:

Attached hereto is a copy of the Thirty Seventh Technical Report Entitled "Effect of Stress on Creep at High Temperatures". This report was prepared under Contract N7-onr-295, Task Order II, NR-031-048. The wholehearted cooperation of the Office of Naval Research in making these studies possible is sincerely appreciated.

Respectfully submitted,

John E. Dorn

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Professor of Physical Metallurgy

JED:dk

EFFECT OF STRESS ON CREEP AT HIGH TEMPERATURES

By

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September 1, 1954

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ABSTRACT

Experimental investigations revealed that the high temperature creep rate, $\dot{\varepsilon}$, is related to the stress, σ, by $\dot{\varepsilon} \sim \sigma^n$ for low stresses and $\dot{\varepsilon} \sim e^{B\sigma}$ for high stresses where n and B are constants independent of the creep strain and temperature. According to a preliminary dislocation climb model for high temperature creep, the activation energy for creep should be that for self-diffusion and the effect of stress on the creep rate should depend on the number of active Frank-Read sources, and the rate of climb depends on the structure as determined by the pattern of climbing dislocations. Many of the experimental observations on high temperature creep can be accounted for by this model.

INTRODUCTION

The creep of metals depends on the two externally adjustable variables of temperature and stress. Recent data[1-5] emphasize that the effect of temperature on the creep rate can be formulated to be

$$\dot{\varepsilon} \sim e^{-\frac{\Delta H}{RT}} \qquad \sigma = \text{constant.} \qquad (1)$$

where $\dot{\varepsilon}$ = creep rate

R = gas constant

ΔH = activation energy for self-diffusion

T = absolute temperature

and σ = stress.

In spite of extensive study few definitive conclusions have yet matured on the effect of stress on the creep rate. Several factors have contributed to the vague opinions currently prevalent on the effect of stress on the creep rate:

1. Theory. The predicted effect of stress on the creep rate was based on unrealistic and immaturely conceived models of the creep phenomenon at high temperatures.

2. Experiment. In general no attempt was made to correct for the fact that different substructures are produced during creep at different stresses. Consequently the observed effect of stress on the creep rate included the unknown effect of differences in structures.

Theories of creep can be generalized into one of three major categories:

1. Creep arises from the continued generation of dislocations. Under these conditions the Becker-Orowan[6-9] type of relationship holds where

$$\dot{\varepsilon} \sim e^{-\frac{N(\sigma_0-\sigma)^2 V}{2GRT}} \qquad (2)$$

and σ_0 = yield strength at $T = 0$

G = shear modulus of elasticity

V = volume of unit under deformation stress σ

N = Avogadro's number.

This relationship, however, is contrary to the observed facts which substantiate Eq. 1 and therefore insist that the temperature and stress terms on the creep rate are separable.

2. Creep arises from thermal activation of dislocations over energy barriers in the slip plane. Under these conditions the Kauzmann[10] type of relationship

$$\dot{\varepsilon} \sim e^{-\frac{\Delta H}{RT}} \sinh \frac{\lambda \sigma}{RT} \tag{3}$$

is obtained where λ depends on the length of the activated dislocation loop. This relationship gives

$$\dot{\varepsilon} \sim e^{-\frac{(\Delta H - \lambda \sigma)}{RT}} \qquad \frac{\lambda \sigma}{RT} >> 1 \tag{4}$$

and

$$\dot{\varepsilon} \sim e^{-\frac{\Delta H}{RT}} \sigma \qquad \frac{\lambda \sigma}{RT} << 1 \tag{5}$$

where ΔH depends on the barrier strengths. The relationship given by Eq. 4 does not agree with experiment because σ is known not to enter the creep relationship as $\frac{\sigma}{T}$.

3. Creep arises from stress relief at a Frank-Read source as a result of dislocation climb to new slip planes. Under these conditions Mott[11] suggests a relationship similar to that given by Eq. 4, where ΔH is now that for self-diffusion. And again this theory fails to agree with all of the facts.

Elaborations of generation or activation theories (1 and 2 above) for creep will not reduce their deviations from the facts. The dislocation climb theory (3 above) appears most promising since it predicts that ΔH should be that for self-diffusion as is observed. It apparently contains a resolvable misconception of how the stress affects the creep rate. But, until this issue is resolved, it too must be considered inadequate to account for the facts.

Examples of experimentally determined effects of stress on the creep rate are summarized in Table I. With the exceptions of the linear relationship of type A, all evaluations were conducted at the secondary creep rate and consequently are influenced by effects of structural differences introduced by the creep straining at different stresses. The linear relationship given by Type A for low stress, however, was obtained under conditions where no initial plastic straining was obtained. Consequently the initial creep rates in these tests refer exclusively to the original annealed structure. Although these data are free from the objections that have been raised against the remaining types, they are not definitive because any stress function whatsoever must become linear over a sufficiently narrow range of the applied stress.

Recent investigations[20] on the effect of stress on the creep rate have been conducted under conditions where the structure was maintained c .stant, indicating that

$$\dot{\varepsilon}\, e^{\frac{\Delta H}{RT}} \sim e^{B\sigma} \qquad (6)$$

In these investigations each specimen was precrept under identical conditions to the same strain whereupon the stress was decreased to a new value

TABLE I

Proposed Relations on Effect of Stress on Creep Rate of Metals

Type	Proposed Relation Between Stress and Creep Rate	Reference	Metal Used, Temperature and Stress Range Investigated: Metal	Temperature (^{o}K)	Stress psi	Remarks
A	Linear Relation $\dot{\varepsilon} \sim \sigma$	Chalmers[12]	Tin	294	14-270	Instantaneous rates on stressing
		Puttick and King[13]	Tin	453-498	7-30	Instantaneous rate of boundary shear of tin bicrystals at various stresses
B	Power Relation $\dot{\varepsilon} \sim \sigma^n$	Buttner, Funk and Udin[14]	Gold	1315	0.7-28	Average over 9 hours
		Alexander, Dawson and Kling[15]	Gold	1193-1293	1.4-218	Secondary creep rate
		Norton[16]	Steel	811-1089	750-25,000	
		Servi and Grant[17]	Pure Al 2S Aluminum 3S Aluminum	366-866 533-866 755	80-6,000 250-6,000 1,000-3,000	Secondary creep rate
C	Exponential Relation $\dot{\varepsilon} \sim e^{B\sigma}$	Dushman, Dunbar and Huthsteiner[18]	Constantan Aluminum Platinum	673-783 523-673 1391-1558	16,000-37,800 360-1450 1450-3630	Secondary creep rate
D	Hyperbolic Sine Relation $\dot{\varepsilon} \sim \sinh B\sigma$	Nadai[19]	Lead			Minimum creep rate

and the initial creep rate immediately following the reduction in stress was measured. Repetitions of this procedure for a series of new decreased stresses permitted correlations of the new initial creep rates as a function of the new stresses. Inasmuch as the precreep conditions were held invariant for any one series of tests, the structure immediately following a decrease in stress was identical for all members of the series. (Increasing the stress was purposely avoided because such higher stresses might have induced immediate straining which in turn would have modified the structure in a way that would have been dependent on the magnitude of the new stress.) Investigations under various precrept conditions revealed that B was insensitive to the stress, strain and temperature conditions of precreep as well as the instantaneous temperature. Solid solution alloying, however, gave lower values of B ; the B values for a cold worked metal[21] were lower than those for the annealed state but they increased during creep and approached those of the annealed state after rather extensive precreep. These results insist on the separation of the temperature sensitive and stress sensitive portions of the creep rate equation.

The dependence of the creep rate on the temperature and stress for a given structure, Eq. 6, might equally be written as

$$\dot{\varepsilon}\, e^{\frac{\Delta H}{RT}} \sim \frac{1}{2} \sinh B\sigma \quad . \tag{7}$$

Then the creep laws over various ranges of stress would reduce to the following limiting relationships:

1. High $B\sigma$ $$\dot{\varepsilon}\, e^{\frac{\Delta H}{RT}} \sim e^{B\sigma} \tag{8}$$

2. Intermediate $B\sigma$ $$\dot{\varepsilon} e^{\frac{\Delta H}{RT}} \sim \frac{1}{2} \sinh B\sigma \quad (9)$$

3. Low $B\sigma$ $$\dot{\varepsilon} e^{\frac{\Delta H}{RT}} \sim \frac{B\sigma}{2} \quad (10)$$

In fact, over sufficiently narrow ranges of stress in the intermediate range of $B\sigma$, a σ^n law might also be satisfactory. If the hyperbolic sine law is valid over the entire stress range, n of σ^n in the intermediate range will vary dependent on the exact range of stresses investigated.

Although reasonably definitive confirmation has been quoted above for the high and low stress laws for creep under identical structures, the best creep rate law for the intermediate stress range has not yet been evaluated. This evaluation is important from a number of aspects:

1. A knowledge of the creep law over the entire range of stresses is of substantial importance to the design engineer.

2. The creep rate law over the intermediate ranges of stress will prove to be definitive in either acceptance or rejection of the suggestion that the creep rate depends on the hyperbolic sine of the stress over the entire range of stress.

3. A definitive evaluation of the creep rate law over the intermediate stress range will assist in the development of more realistic theories of creep. These factors stimulated the following investigations

MATERIALS AND TECHNIQUES

The series of alpha solid solutions of magnesium in high purity aluminum identified in Table II were used in the present investigation. After machining, the specimens were annealed as described in Table II to achieve about the same grain size.

All creep tests were conducted under constant load conditions, the temperature being held constant to within $\pm 1^\circ C$ of the reported values.

TABLE II

Chemical Analyses, Annealing Treatment and Grain Size of Aluminum Alloys Investigated

Atomic Percent Mg	Chemical Analyses (Wt. % Impurities)					Annealing Treatment of As Received Stock	Grain Diam. (cm)
	Cu	Fe	Si	Mn	Mg		
0	.001	.000	.001	.000	.001	50 mins at 510°C	.08
1.06	.001	.001	.001	.001	--	65 mins at 540°C	.08
2.09	.001	.002	.001	.001	--	65 mins at 540°C	.08
3.12	.001	.002	.001	.001	--	60 mins at 600°C	.08

The stress was measured to within ± 20 psi and the creep strains to within ± 0.0001.

EXPERIMENTAL RESULTS

At sufficiently low creep stresses, the deformation is exclusively elastic and the initial plastic strain is zero. Typical examples of creep curves for such cases are given in Fig. 1 where the elastic component of the strain has been subtracted from the total strain to give a true creep strain. Consequently the initial creep rates refer exclusively to the effect of stress on the creep rate since the initial structure is always that for the previously annealed condition. As shown in Fig. 2 the initial creep rate is related to the stress according to

$$\dot{\varepsilon} \sim \sigma^{n} \tag{11}$$

where n (the slope of the log $\dot{\varepsilon}$ versus log σ curve) is insensitive to the absolute temperature. This again suggests that the stress and temperature terms for high temperature creep are separable, as shown in Fig. 3, where

$$\dot{\varepsilon} e^{\frac{\Delta H}{RT}} \sim \sigma^{n} \tag{12}$$

over the range of temperatures from 531° to 853°K. The fact that ΔH is 36,000 cal/mole in the low stress tests as well as for the previously reported high stress tests[1], further emphasizes the fact that the activation energy for high temperature creep is insensitive to the stress.

Similar tests on the various alpha solid solutions of Mg in Al identified in Table II are correlated in Fig. 4. Thus, within the normal scatter

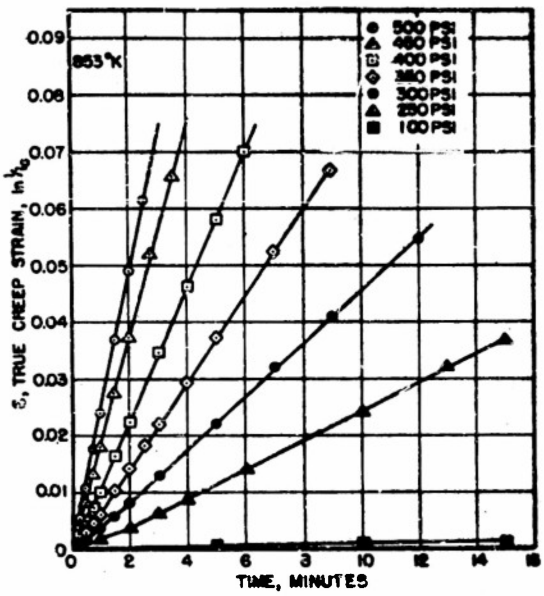

FIG. I CONSTANT LOAD CREEP OF ALUMINUM CONTAINING 3.I ATOMIC PERCENT MAGNESIUM AT LOW STRESSES.

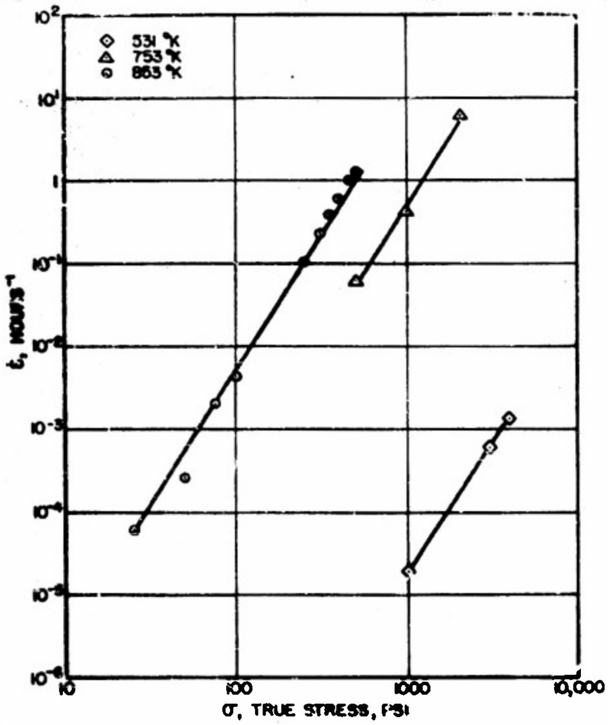

FIG. 2 EFFECT OF SMALL STRESSES ON THE INITIAL CREEP RATE OF ALUMINUM CONTAINING 3.I ATOMIC PERCENT MAGNESIUM.

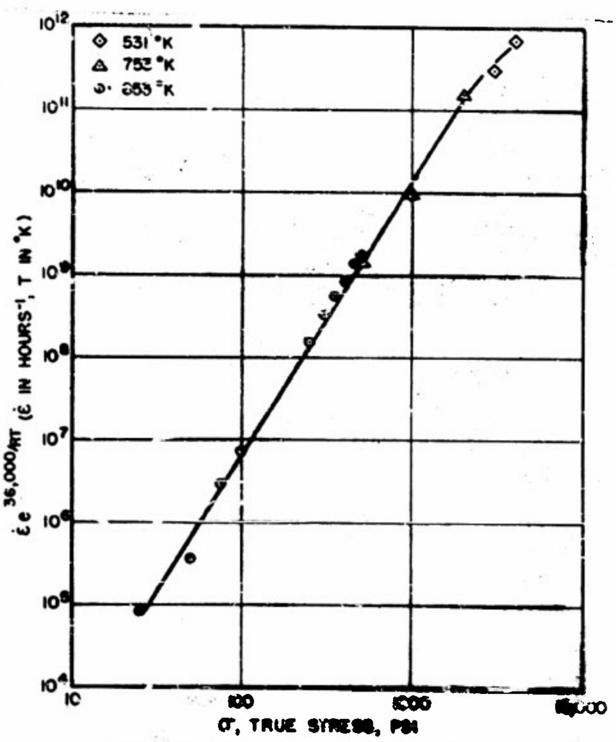

FIG. 3 EFFECT OF SMALL STRESSES ON THE INITIAL CREEP RATE OF ALUMINUM CONTAINING 3.I ATOMIC PERCENT MAGNESIUM.

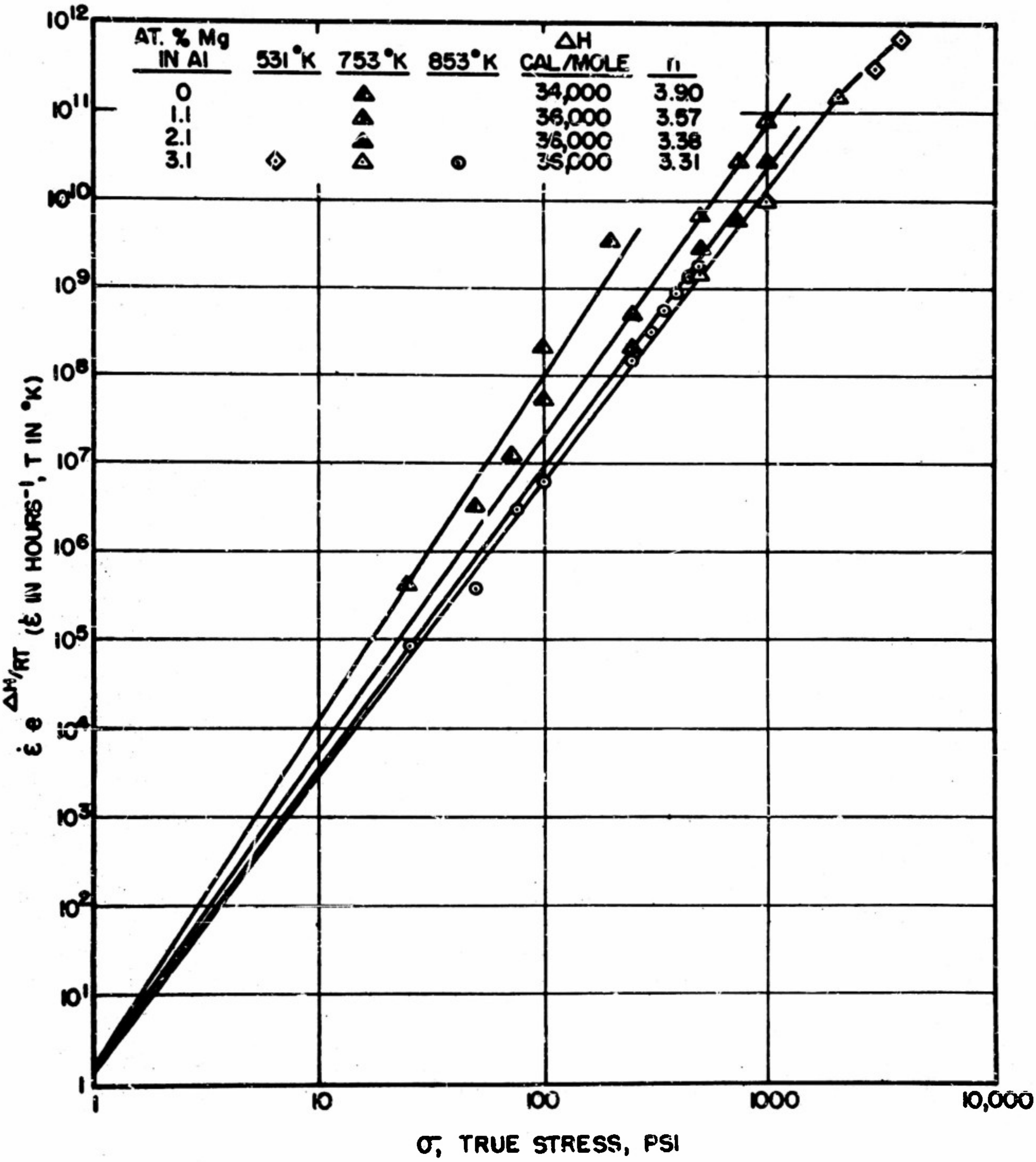

FIG. 4 EFFECT OF SOLID SOLUTION ALLOYING ON THE INITIAL CREEP RATE AT LOW STRESSES.

in the data,

$$\dot{\varepsilon} = S' e^{-\frac{\Delta H}{RT}} \sigma^n \tag{13}$$

where n decreases with alloying and the parameter S' (obtained by extrapolating $\dot{\varepsilon} e^{\frac{\Delta H}{RT}}$ to unit stress) for the annealed state is insensitive to alloying.

Therefore these data suggest that the creep law for high temperatures is

$$\dot{\varepsilon} = S e^{-\frac{\Delta H}{RT}} \varphi(\sigma) \tag{14}$$

where $S\varphi(\sigma) = S'\sigma^n$ for intermediate stresses and $S\varphi(\sigma) = S'' e^{B\sigma}$ for high stresses. Unfortunately the previously reported high stress tests[20,21] did not extend to sufficiently low values to include the σ^n range and the intermediate stress tests reported above could not be extended to sufficiently high stresses to enter the $e^{B\sigma}$ range because of introduction of the complicating factor of initial plastic straining. Thus the merging of the two ranges of conditions was not verified under a single test procedure but implied by two sets of results obtained under alternate test conditions. In order to provide unambiguous evidence of the merging of the two ranges of conditions, the additional tests documented in Fig. 5 were undertaken. Here again the previously described decrease in stress technique was adopted in order to maintain the same structures for any one sequence of tests. Again the $e^{B\sigma}$ relationship was observed to be valid over the high stress range. But consistent deviations from the $e^{B\sigma}$ relationship were obtained, as shown by the data of Fig. 5, at appropriately low stresses. As indicated in Fig. 6 the creep rate at these low stresses

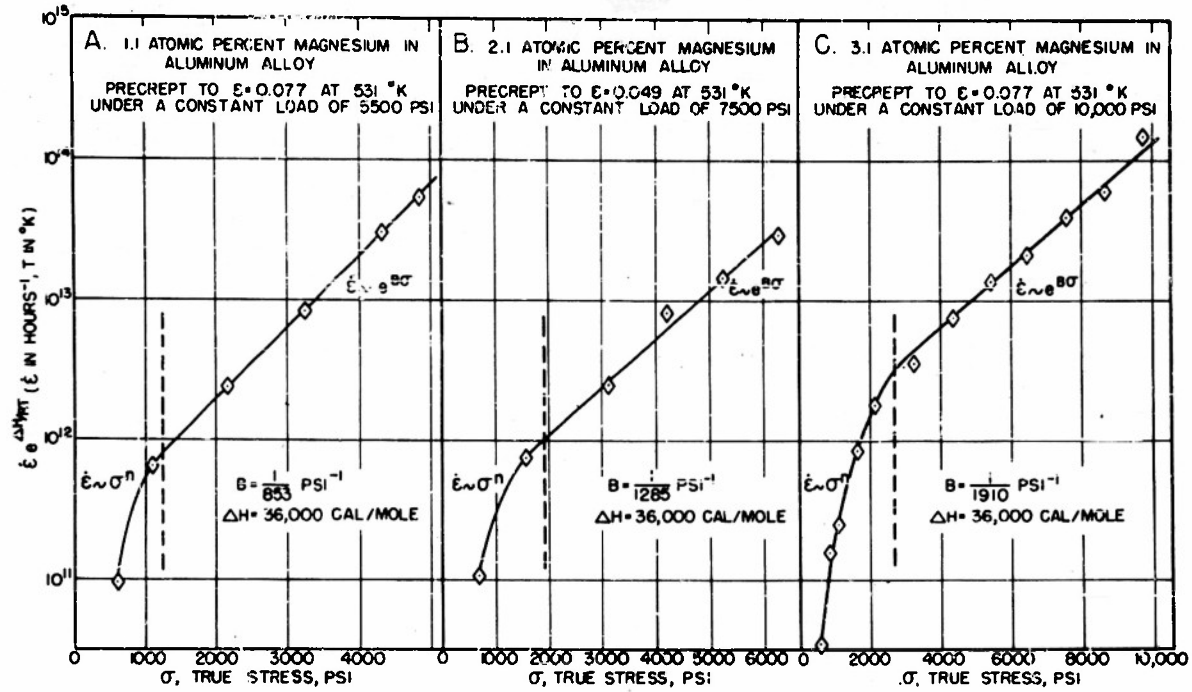

FIG. 5 EFFECT OF STRESS ON THE CREEP RATE AT A GIVEN STRUCTURE.

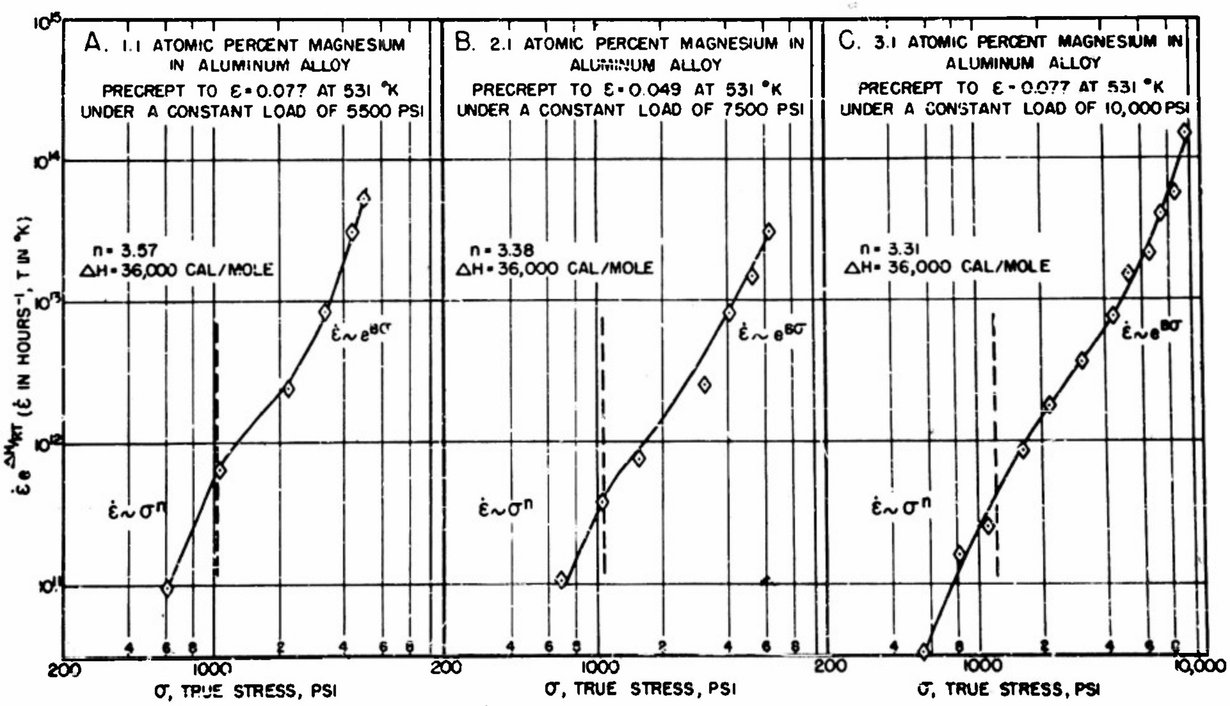

FIG. 6 EFFECT OF STRESS ON THE CREEP RATE AT A GIVEN STRUCTURE.

becomes proportional to σ^n. Thus it appears possible to go from σ^n to the $e^{B\sigma}$ law in one series of tests.

If the two ranges of creep laws merely represent ranges of transition of a single function from $e^{B\sigma}$ to σ^n, the transition must occur at the same values of $B\sigma$ and σ^n independent of the values of B and n. The data given in Fig. 7 reveals that the transition occurs at $B\sigma \cong 1.5$ whereas the data of Fig. 8 shows that the transition occurs at $\sigma^n \cong 10^{11}$. The different elevations of the curves are ascribable to the different values of the structure sensitive parameters S' or S''.

Therefore the validity of Eq. 14 for the creep rate is quite well established. Careful examination of the data, however, clearly demonstrates that $\varphi(\sigma)$ is not the hyperbolic sine function. And thus far no single function has been found that will agree well with the data over the entire range of stresses employed, in spite of the proof given above that such a function exists.

Up to the present only the initial creep rates following either the application of a low stress or the reduction of a precreep stress were considered. In both cases, however, transients were observed which are believed to be important in the formulation of an accurate theory for high temperature creep. Upon initial loading to low stresses transients of the type shown in Fig. 9A were obtained, where with continued time at the stress the creep rate increased before creep entered the usual range of the primary stage of decreasing creep rates. Upon decrease of the stress the initial creep rate was always higher than those obtained after a short period of additional creep as shown in Fig. 9B. Such transients persisted only for small strains.

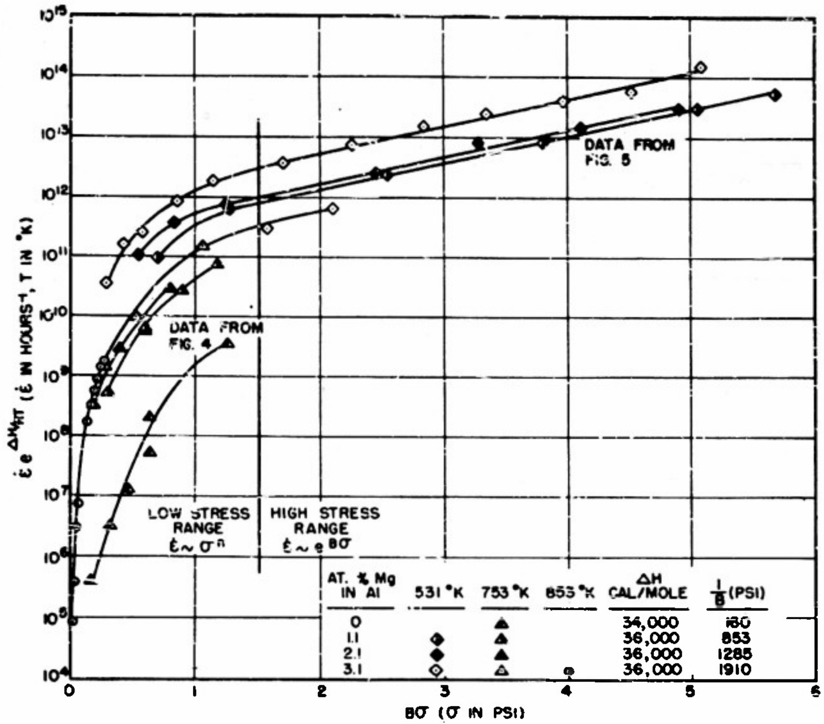

FIG. 7 CORRELATION OF THE DATA IN FIGURES 4 AND 5 IN TERMS OF $B\sigma$.

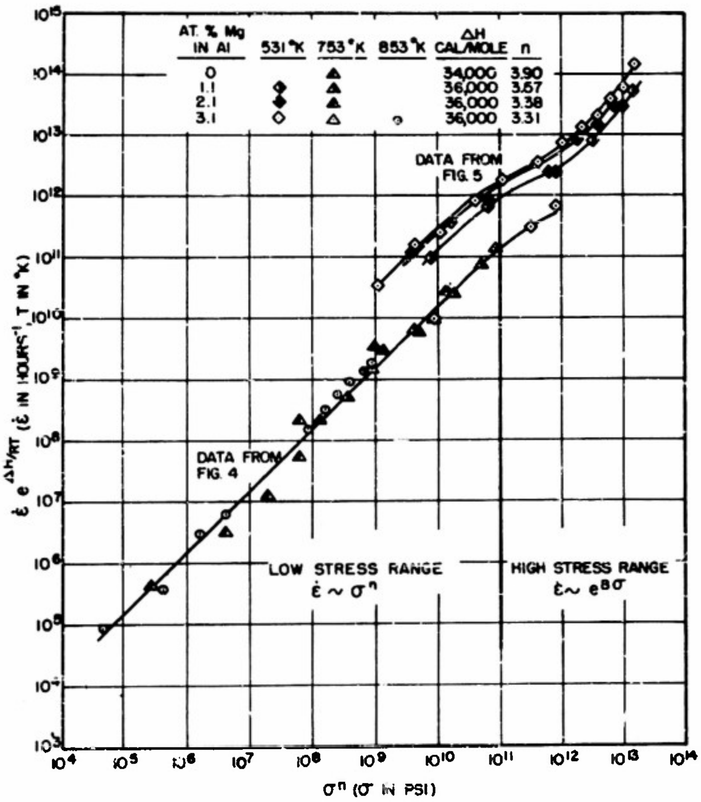

FIG. 8 CORRELATION OF THE DATA IN FIGURES 4 AND 5 IN TERMS OF σ^n.

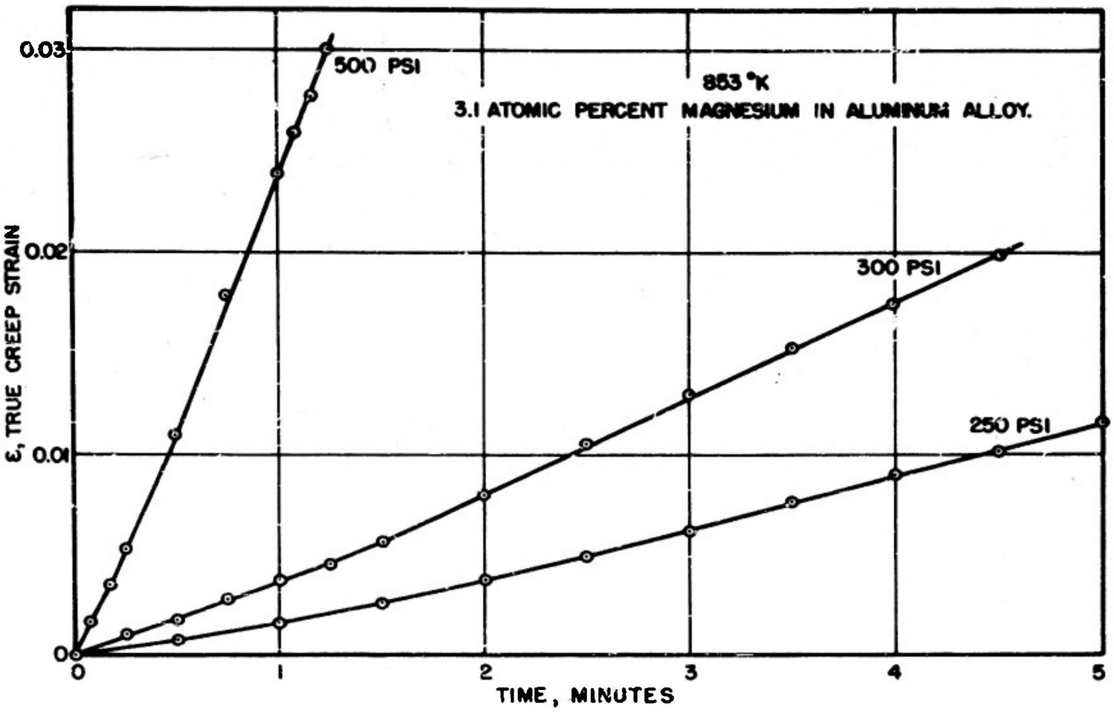

FIG. 9A TRANSIENTS UPON APPLYING LOW STRESSES.

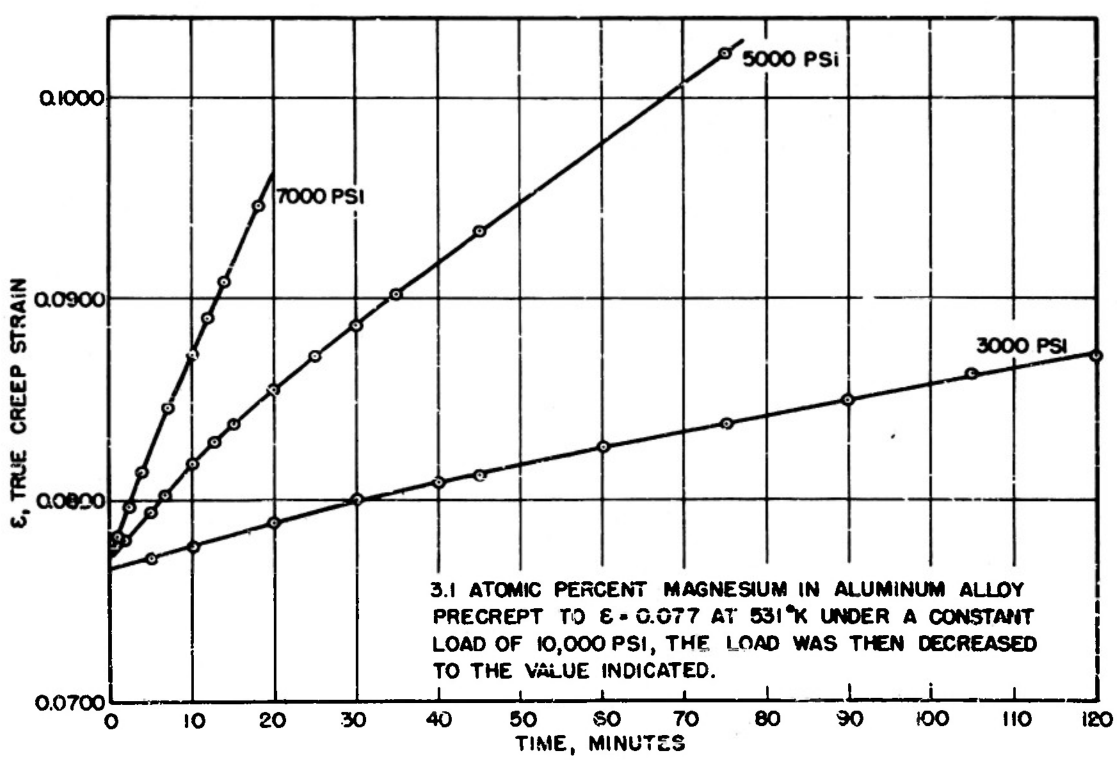

FIG. 9B TRANSIENTS UPON DECREASING THE STRESS.

Another abnormality was observed in the decrease in stress type of data as shown in Fig. 10. The precreep datum just before the decrease in stress is given by the ◆ symbol. Not until the stress was decreased more than about 2000 psi below the precreep stress was the $e^{B\sigma}$ relationship obtained. Similar observations were made in all tests undertaken and the deviations of the final precreep datum from the extrapolated $e^{B\sigma}$ relationship was found to increase with increasing precreep stress.

DISCUSSION

The identity of the activation energy for high temperature creep with that for self-diffusion strongly suggests that high temperature creep is controlled by self-diffusion. Inasmuch as stress directed self-diffusion alone cannot account for the observed creep rates, it is probable that creep occurs by a dislocation climb process wherein the rate of self-diffusion determines, in part, the rate of climb. The insensitivity of the activation energy term to the stress and the consequent separation of the temperature and stress effects into two independent multiplicative terms might at first appear unorthodox.

In general the free energy for activation of a positive unit climb of a single dislocation will be $\Delta h - T\Delta s - \left(-\frac{dU}{dy}a\right)$ where Δh is the activation energy for self-diffusion, Δs is the entropy of activation, $\left(-\frac{dU}{dy}\right)$ is the potential energy gradient in the direction of climb and a is the atomic climb spacing. A similar expression applies for a negative climb process excepting that the sign preceding the potential energy term is changed because negative climb processes increase the energy of the system. Thus the net frequency of a unit climb is approximated by

$$\nu = 2\frac{kT}{h} e^{\frac{\Delta s}{k}} e^{-\frac{\Delta h}{kT}} \sinh\left\{\left(-\frac{dU}{dy}\right)\frac{a}{kT}\right\}$$

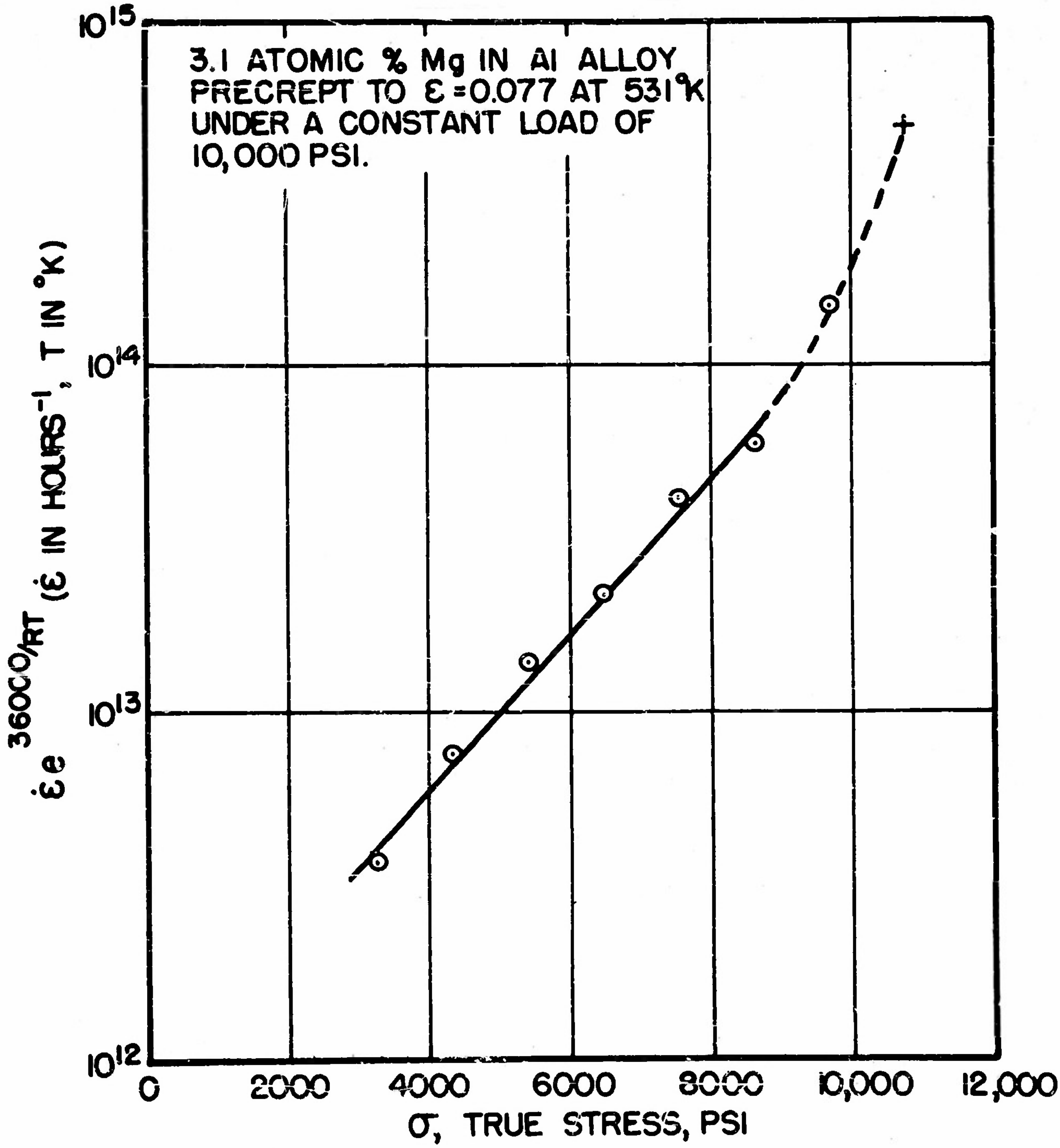

FIG. 10 FAILURE OF $e^{B\sigma}$ RELATIONSHIP FOR SMALL DECREASES IN STRESS.

Mott's analysis[11] suggests that $\left(-\frac{dU}{dy}\right)$ is linearly dependent on the stress. But a simple calculation will reveal that $\left(-\frac{dU}{dy}\right) = 0$ when the dislocations from a single Frank-Read source are yet on their original slip plane. In general a shear stress in the direction of slip cannot promote climb. In fact in order to have a non-vanishing value of $\left(-\frac{dU}{dy}\right)$ at a given dislocation it is necessary that surrounding dislocations be on other slip planes. Thus the value of $\left(-\frac{dU}{dy}\right)$ that directs the climb of any one dislocation depends on the surrounding pattern of dislocations or, in other words, the structure. This structure is dependent primarily on the pre-creep history and is evidently insensitive to the instantaneous applied stress. The experimental evidence shows that $\left(-\frac{dU}{dy}\, a\right)$ must be small since the total value of $\left(\Delta h + \frac{dU}{dy}\, a\right)$ is known to be almost that for self-diffusion, namely Δh alone. If this is so, the hyperbolic sine term reduces to its argument and

$$\nu = \frac{2a}{h}\left(-\frac{dU}{dy}\right) e^{\frac{\Delta s}{k}}\; e^{-\frac{\Delta h}{kT}}$$

where the exponential term now gives only the Δh term for self-diffusion.

Under the assumptions made in this dislocation climb model for high temperature creep, straining takes place because the climbing dislocations result in a reduction of the back stress at a Frank-Read source and thus allow the generation of new dislocations. The actual details of this process are so complex that for the present they defy an accurate mechanistic analysis. But some concept of this process might be gained from a gross statistical approach to the problem. Let N_c be the number of dislocations that have been generated by a single source. Let A be the average area swept out by each of these dislocations on their slip planes when a single dislocation undertakes a unit climb. If N are the number of sources per

unit volume the creep strain per unit climb is

$$\varepsilon = A N N_c b$$

and the creep rate becomes

$$\dot{\varepsilon} = 2 A N N_c b \frac{a}{h} \left(-\frac{dU}{dy}\right) e^{\frac{\Delta s}{k}} e^{-\frac{\Delta h}{kT}} \tag{15}$$

In general therefore A , N_c and $\left(-\frac{dU}{dy}\right)$ must be given appropriate average values. Like $\left(-\frac{dU}{dy}\right)$ the values of A and N_c also depend on the instantaneous structure.

The number of active Frank-Read sources N depends on the applied stress and the distribution of source lengths. The stress necessary to promote creep at a Frank-Read source[20] will be estimated to be

$$\sigma = \frac{Gb}{L} \tag{16}$$

where G = shear modulus

b = Burgers vector

L = source length.

The number of sources having lengths between L and $L + dL$ can be represented by

$$dN = \psi'(L)\, dL \tag{17}$$

All sources having lengths greater than $L_c = \frac{Gb}{\sigma}$ will be active under stress σ . And all such sources will remain active because the back stresses on these sources will be continually relieved by the climb process. Thus no exhaustion of sources takes place and

$$N = \int_{L_c}^{\infty} \psi'(L)\, dL = \psi(L_c) = \varphi\left(\frac{\sigma}{Gb}\right) \tag{18}$$

Consequently the creep rate is given by

$$\dot{\varepsilon} = 2AN_cb\left(-\frac{dU}{dy}\right)\frac{a}{h}e^{\frac{\Delta s}{k}}e^{-\frac{\Delta h}{kT}}\varphi\left(\frac{\sigma}{Gb}\right) \tag{19}$$

For a given structure A , N_c and $\left(-\frac{dU}{dy}\right)$ are fixed quantities and Eq. 19 reduces to the experimentally verified law of Eq. 14.

In addition to accounting for the observed dependence of the creep rate on stress and temperature, the theory exhibits other virtues. But in view of the simplifying averaging methods that were employed in lieu of a detailed mechanistic model of release of new dislocations from Frank-Read sources, the theory cannot yet be expected to account for all of the observed facts.

1. Primary Creep. According to the theory primary creep is due to the decrease in the structure dependent term $AN_c\left(-\frac{dU}{dy}\right)$. But it is not immediately apparent that this product should decrease over the primary range.

2. Secondary Creep. As creep continues under a given stress, the dislocations will climb to subboundaries. Finally a steady state pattern of dislocations will be obtained providing a basis for secondary creep.

3. Tertiary Creep. Tertiary creep is extraneous to the simple dislocation climb theory since it probably arises from condensation of vacancies and consequent growth of microcracks. It should be possible to introduce the essential features leading to tertiary creep and microfracturing in the current model.

4. Mechanical Equation of State. According to the dislocation climb model each stress produces its unique structure. At high stresses the number of climbing columns of dislocations is greater due to the greater initial packing of dislocations along the slip plane. Thus the structure

developed during a dislocation climb process will be stress dependent (and insensitive to the temperature) as well as strain dependent in conformity with experimental observations. For this reason the creep rate becomes a function of the past stress history as well as the instantaneous conditions of test. Necessarily the mechanical equation of state fails to apply to creep.[2]

5. Transients. Upon first loading $\left(-\frac{dU}{dy}\right)$ is small but as soon as a few dislocations climb above their original slip planes it should increase due to the then greater increase in the potential energy gradient in the climb direction. This is in conformity with the observed transient upon applying low stresses.

Immediately upon decreasing the stress there exists a greater number of climbing dislocations than the steady state number for that stress. Consequently the initial creep rate upon a decrease in stress is greater than the steady state value.

6. Effects of Alloying. The preliminary estimate given by Eq. 16 for the stress necessary to activate a Frank-Read source neglected the effect of non-conservative losses. Thermal lattice vibrations and localized strain regions about solute atoms are sources of internal stresses that react with the moving dislocation. Since the energy of a system consisting of a single constant length straight dislocation is unmodified by its position in the crystal the energy required to move the dislocation through the stress field is dissipated thermally. Therefore the actual stress required to activate a source is somewhat larger than that given by Eq. 16, say

$$\sigma = \left(\frac{Gb}{L}\right)^{1+k}$$

where k is a small quantity. Thus the function φ of Eq. 19 takes the form

$$\frac{\sigma^{\frac{1}{1+k}}}{Gb}$$

As suggested previously k increases with alloying. Consequently n of Eq. 14 should decrease with alloying.

CONCLUSIONS

1. The activation energy for high temperature creep of aluminum and its dilute alloys is insensitive to the variables of stress and strain and approximates the value for self-diffusion.

2. The creep rate equation for a constant structure is approximated by

$$\dot{\varepsilon} = S e^{-\frac{\Delta H}{RT}} \varphi(\sigma)$$

where $\dot{\varepsilon}$ = creep rate

S = parameter that depends on the structure

ΔH = activation energy for self-diffusion

R = gas constant

T = absolute temperature

σ = stress

$$S\varphi(\sigma) = S'\sigma^{n} \qquad \sigma^{n} \lessdot 10^{11} \quad \text{or} \quad B\sigma \lessdot 1.5$$

$$S\varphi(\sigma) = S''e^{B\sigma} \qquad \sigma^{n} \gtrdot 10^{11} \quad \text{or} \quad B\sigma \gtrdot 1.5$$

The creep rate does not appear to be a hyperbolic sine function of the stress.

3. Most of the available experimental evidence strongly supports the dislocation climb model for high temperature creep.

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REFERENCES

1. O. D. Sherby and J. E. Dorn, "Creep Correlations in Alpha Solid Solutions of Aluminum", Transactions, AIME, vol. 194, 1952, pp. 959-964.

2. O. D. Sherby and J. E. Dorn, "Some Observations on Correlations between the Creep Behavior and the Resulting Structures in Alpha Solid Solutions", Transactions, AIME, vol. 197, 1953, pp. 324-330.

3. O. D. Sherby, R. L. Orr and J. E. Dorn, "Creep Correlations of Metals at Elevated Temperatures, Journal of Metals, AIME, January 1954, pp. 71-80.

4. R. L. Orr, O. D. Sherby and J. E. Dorn, "Correlations of Rupture Data for Metals at Elevated Temperatures", Transactions, ASM, vol. 46, 1954, pp. 113-128.

5. B. Fazan, O. D. Sherby and J. E. Dorn, "Some Observations on Grain Boundary Shearing During Creep", Journal of Metals, AIME, August 1954, pp. 919-922.

6. R. Becker, "Über die Plastizität amorpher und kristalliner fester Körper", Physikalische Zeitschrift, vol. 26, 1925, pp. 919-925.

7. R. Becker, "Über Plastizität, Verfestigung und Rekrystallization", Zeitschrift fur Tech. Physik, vol. 7, 1926, p. 547-555.

8. E. Orowan, "Crystal Plasticity I, II, III", Zeitschrift fur Physik, vol. 89, 1934, pp. 605-659.

9. E. Orowan, "The Creep of Metals", Journal, West of Scotland Iron and Steel Institute, February 1947.

10. W. Kauzman, "Flow of Solid Metals from the Standpoint of the Chemical-Rate Theory", Transactions, AIME, vol. 143, 1941, pp. 57-83.

11. N. F. Mott, "A Theory of Work Hardening of Metals, II. Flow without Slip Lines, Recovery and Creep", Philosophical Magazine, vol. 44, Series 7, July 1953, pp. 742-765.

12. B. Chalmers, "Microplasticity in Crystals of Tin", Proceedings, Royal Society, London, vol. 156, 1936, pp. 427-443.

13. K. E. Puttick and R. King, "Boundary Slip in Bicrystals of Tin", Journal, Institute of Metals, 1952, vol. 80, pp. 537-544.

14. F. H. Buttner, E. R. Funk and H. Udin, "Viscous Creep of Gold Wires Near the Melting Point", Transactions, AIME, vol. 194, 1952, pp. 401-407.

15. B. M. Alexander, M. H. Dawson and H. P. Kling, "The Deformation of Gold Wire at Elevated Temperatures", Journal of Applied Physics, vol. 22, 1951, pp. 439-443.

16. F. H. Norton, "Creep of Steel at High Temperatures", McGraw Hill, New York, 1929.

17. I. S. Servi and N. J. Grant, "Creep and Stress-Rupture Behavior of Aluminum as a Function of Purity", Transactions, AIME, 1951, vol. 191, pp. 909-916.

18. S. Dushman, L. W. Dunbar and H. H. Huthsteiner, "Creep of Metals", Journal of Applied Physics, vol. 15, 1944, pp. 108-124.

19. A. Nadai, "The Influence of Time Upon Creep. The Hyperbolic Sine Creep Law", Stephen Timoshenko Anniversary Volume, MacMillan, New York, 1938, pp. 155-170.

20. O. D. Sherby, R. Frenkel, J. Nadeau and J. E. Dorn, "Effect of Stress on the Creep Rates of polycrystalline Aluminum Alloys under Constant Structure", Journal of Metals, AIME, February 1954, pp. 275-279.

21. R. E. Frenkel, O. D. Sherby and J. E. Dorn, "Effect of Cold Work on the High Temperature Creep Properties of Dilute Aluminum Alloys", to be published in Transactions, ASM, 1955.

22. F. C. Frank and W. T. Read, Jr., "Multiplication Forces for Slow Moving Dislocations", Physical Review, 1950, vol. 79, pp. 722-723.

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